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RESIDUAL STRESS DEVELOPMENT DURING FABRICATION AND PROCESSING OF GAMMA-TITANIUM BASED COMPOSITES

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Introduction

Titanium aluminide-based intermetallic matrix composites (IMCs) have received considerable attention as potential replacements for current high-temperature structural materials. However, one of the major hindrances to their use is the development of high residual stresses during processing which can result in localized plastic deformation, matrix cracking, and ultimately in premature failure [1,2]. Titanium aluminide based on the γ-Ti are considered as low-density, high-strength high temperature materials used in the hot sections of jet engines [3,4]. Like most intermetallics, γ-Ti alloy suffer from low ductility and fracture toughness at ambient temperature and low oxidation resistance at temperatures significantly lower than its melting temperature. The use of an alloying addition such as manganese for ductility and niobium for increased oxidation resistance can improve these properties while maintaining the high modulus and creep resistance of the gamma system [5]. Residual stresses in composites are typically thermal in nature and are the result of the difference in coefficients of thermal expansion (CTE) between the fibers and the matrix. Composites are typically produced at elevated temperatures where both the fiber and the matrix are essentially stress free. During cooling to ambient temperatures, however, residual stresses arise due to the thermomechanical (*i.e.*, CTE, strength, etc.) mismatch between the fibers and the matrix. Though it is not possible to eliminate the residual stresses that develop during processing, it is possible to minimize their influence by carefully controlling the processing path and the temperature cycles.

In this paper, the result of a study on the influence of various thermal treatments on residual stresses in γ-TiAl-based IMCs is reported. To accomplish this objective, composite materials were fabricated by hot isostatically pressing (HIP'ing) powdered matrix material and reinforcing fibers. The composite samples were manufactures such that the inter-fiber spacing were at least 5 times the fiber diameter to reduce the fiber to fiber effects. The volume fraction was approximately 5–10%. Alumina and Sapphikon in the form of fiber reinforcement is thermodynamically stable in γ-Ti, and its coefficient of thermal expansion closely matches TiAl [6]. The resulting composites were then heat treated for various combinations of time-temperature combinations designed to simulate important processing and

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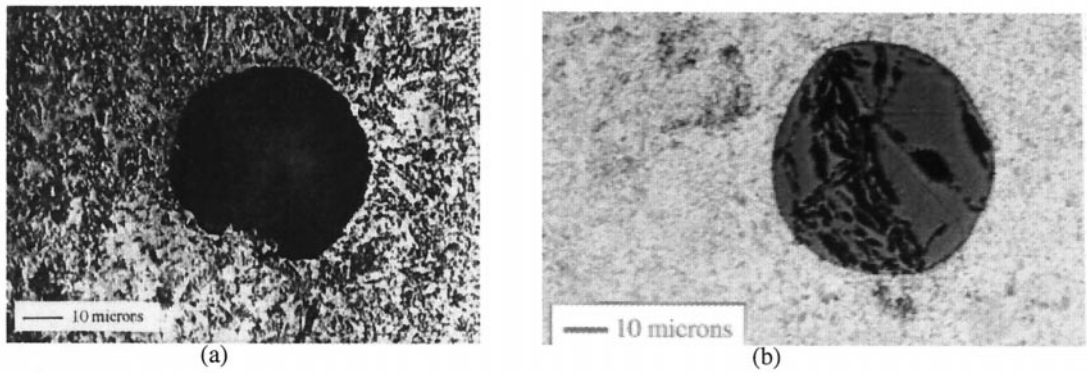

Figure 1. An optical micrograph of the Ti-47–2Ta composite reinforced with a) alumina fiber b) Sapphikon fiber.

heat treatment conditions. Residual stresses were evaluated in the intermetallic matrix using combinations x-ray diffraction (XRD). A minimum residual stress was discovered for each fiber at an optimum temperature and heat treatment. A composite cylinder model was developed and the predictions showed good comparison to the experimental results. The results show that the relaxation and increase in the residual stress data can be explained based on the creep of the intermetallic matrix at the heat treatment temperatures.

Material and Experimental

Ti-47Al-2Ta matrix composites reinforced with alumina and Sapphikon were employed in this investigation (Fig. 1). Both the alumina polycrystalline fiber and Sapphikon fibers had an average diameter of 50 μm. The composite material was fabricated at the McDonnell-Douglas Research Center using hot isostatic pressing (HIP'ing) of powdered matrix material over the different fiber types. Pre-alloyed powders were milled in SPX800 mills [7]. All powder handling and milling operations were carried out in argon glove boxes and Titanium lined vials to minimize contamination. Due to the larger quantities of powder milled in SPX800, this material was sealed in TI-3AL-2.5V(wt%) tube 6.35 mm in diameter, and were then Hipped at 100°C° and 200 MPa for two hours. Some powder was retained for characterization. The fibers used were purchased from 3M. A total of 15–20 fibers were used in this study to assure a very low volume fraction of fibers (less than 10%). This was necessary to eliminate the influence of fiber to fiber interaction. To create the composite, a thin titanium tube approximately 6.35 mm in diameter was electron beam welded at one end. The selected fibers were dropped in through the open end and the remaining space filled with a powder of the matrix material. When full, the other end was electron beam welded and the material was hot isostatically pressed in vacuum and at 1200°C. The tubes were sectioned into pieces and each piece was encapsulated in quartz for the heat treatment. The composites were heat treated at 593°C, 815°C, or 982°C for 100, 200, or 500 hours followed by cooling to room temperature.

The most widely used nondestructive technique for measuring residual stress is XRD [4–11]. The fundamental equation of x-ray strain determination is [5]:

$$(\epsilon'_{33})_{\phi\psi} = \frac{d_{\phi\psi} - d_0}{d_0} = \epsilon_{11}\cos^2\phi\sin^2\psi + \epsilon_{12}\sin 2\phi\,\sin^2\psi + \epsilon_{22}\sin^2\phi\,\sin^2\psi$$

$$\epsilon_{33}\cos^2\psi + \epsilon_{13}\cos\phi\,\sin 2\psi + \epsilon_{23}\sin\phi\,\sin 2\psi \qquad (1)$$

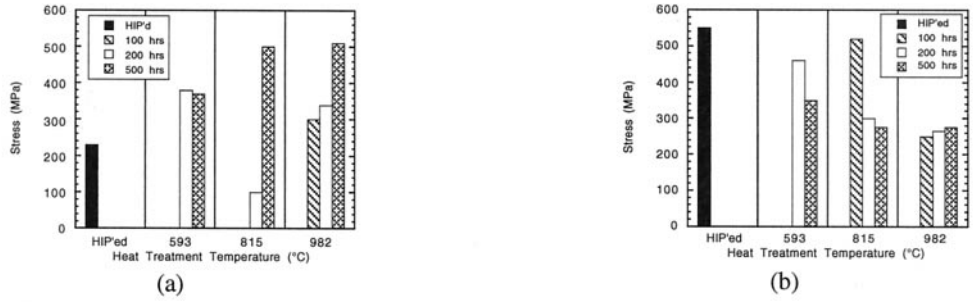

Figure 2. Von Mises Stress following various heat treatments cycles. A) Alumina fibers, b) Sapphikon fiber.

In this equation, ϕ is the material rotation angle, ψ is the material tilt angle, $d^{\phi\psi}$ is the inter planar spacing at angles of ϕ and ψ in the stressed material, d_o is the inter planar spacing of the unstressed material, and ϵ_{ij} are the strain components. The subscripts 1 and 2 denote strains in the surface plane, while 3 indicates strains perpendicular to the surface plane. There are six unknowns in eq. (1) that can all be solved by measuring $d,^{\phi\psi}$ along all six independent directions. The resulting stresses can then be determined using Hooke's law and the von Mises stress is calculated from the stress components [11].

XRD patterns were obtained using a Phillips PW3040 type diffractometer with a PW3050/θ-2θ type goniometer operating at 40 kV and 55 mA. The peak located at a 2θ of 79.3° was used for the stress measurement. To determine all components of stress, nine angles of ψ (0°, ±15.34°, ±21.97°, ±27.34°, ±32.01°) were measured at each of three angles of ϕ (0°, 45°, 90°). The d-spacing was measured at all 27 points for each specimen. The resulting data was then substituted into eqs (1) - (3), to calculate the residual stress for all the stress components and finally, the von Mises stress, which was used as an average/effective measure of stress in the composite. The error based on the errors within the 2theta angle measurement is calculated to be ~20 MPa.

Results

XRD Results

The von Mises stress for different temperatures and heat treatment conditions are calculated from the individual components of stress in the 3-D x-ray stress analysis (Fig. 2). It is clear that the von Mises stress in the alumina reinforced composite increased above that observed in the HIP'ed specimen following the 593°C heat treatment and decreased only slightly with an increase in aging time (Fig. 2-a). This decrease may be interpreted as within the error range of the experimental data. The specimen heat treated at 815°C shows a decrease in von Mises stress from the HIP'ed conditions after 200 hours followed by a large increase in stress after 500 hours. The heat treatment at 982°C results in an increase in stress over the HIP'ed condition and continues to increase after each increment of time. After 100 hours at 982°C, the stress becomes larger than the stress in the HIP'ed specimen. A larger Von Mises stress was measured for the 200 hour time of heat treatment at 593°C, compared to the as-HIP'ed-material. This, however, decreases sharply at 815°C, and then increases to a high value at 982°C. The heat treatments at 500 hours caused the stress to increase as the temperature increased.

The von Mises stress in the Sapphikon reinforced composite, (Fig. 2-b), shows it's highest value in the untreated specimen. The stress steadily decreased as the time of heat treatment increased for both 593°C and 815°C. At a temperature of 982°C, the residual stress resulted in a drop from the HIP'ed value after 100 hours, and then increased slightly with an increase in time. After 100 hours of heat

treatment, the von Mises stress decreased slightly at 815°C and then an increase was shown at 982°C. The increase in time of heat treatment resulted in a decrease in residual stress up to a transitional temperature beyond which the stress increased as the time of heat treatment was increased. This interesting behavior is evident for both Alumina and Sapphikon.

Inelastic Concentric Cylinder Model

A composite cylinder model was chosen for the analysis of the evolution of the residual stresses because of its ability to predict effective composite properties. The model assumes an elastic cylindrical fiber embedded in an inelastically deforming matrix, similar to the models presented by McLean [15] and Fabeny [16]. The effect of other fibers from the surrounding is assumed to be negligible. As a result the model is only valid for the case of composites with relatively low volume fractions. Other researchers [12–14] have used various forms of the model. Plane strain assumption ($\epsilon_z = 0$) together with the Tresca yield criterion (for $\sigma_\theta > \sigma_z > \sigma_r$) and the associated flow rule is used in a cylindrical coordinate system to arrive at a closed form solution. The analysis assumes a uniform temperature throughout the composite. Additionally, it is assumed that body forces and surface tractions are absent [17].

During plastic deformation the general form of Norton's law for multidimensional analysis can be expressed in the form

$$\frac{d\epsilon}{dt} = K_c \sigma^n \, exp\left(\frac{-Q}{RT}\right) \tag{2}$$

Where *kc* is the creep rate parameter (hr^{-1}), *n* is the creep exponent, *Q* is the activation energy (kJ/mol), *R* is the gas constant and *T* is the absolute temperature (K).

Assuming a stress-free outer surface, and a prescribed displacement and radial stress at the interface, as imposed by the fiber and matrix thermal coefficients of expansion α_f and α_m, respectively. The solution to the equilibrium equations and using an elastic-inelastic medium (eq. 1) for the matrix the solution to the stresses (residual stresses) can be approximated by the following expressions

$$S_r(\rho) = \frac{-T_0}{2(1-\nu)}\left(\frac{\rho_0}{\rho}\right)^2 - \frac{B''}{2n} S_0^n \left(\frac{\rho_0}{\rho}\right)^{2n} \tag{3}$$

$$S_\theta(\rho) = \frac{-T_0}{2(1-\nu)}\left(\frac{\rho_0}{\rho}\right)^2 + \left(1 - \frac{1}{2}\right) B'' S_0^n \left(\frac{\rho_0}{\rho}\right)^{2n} \tag{4}$$

where, B, B′ and B″ are found from the following equations:

$$B = K_c(T)\, \exp(Q/RT), \quad B' = B\Delta t, \quad B'' = \frac{2B'\Delta t}{\nu(1+\nu)} \tag{5}$$

ν is the Poisson's ratio and Δt is the time increment, K_c depends on temperature (Table 1). Other dimensionless quantities are defined as:

$$S_r = \frac{\sigma_r}{\sigma^0}, \quad S_\theta = \frac{\sigma_\theta}{\sigma^0}, \quad S_z = \frac{\sigma_z}{\sigma^0}, \quad \tau = \frac{E\alpha T}{(1-\nu)\sigma^0}, \quad \rho = \frac{r}{b}, \quad \beta = \frac{b}{a} \tag{6}$$

a and *b* are the internal and external radii of the concentric cylinder, σ^o is the yield stress, ϵ^o is the yield strain for the matrix and σ^e is the initial elastic stress. A summary of the materials parameters used for the simulation is provided in Table 1.

TABLE 1

Quantity	Symbol	Value
Outside Diameter of Cylinder Model	b	200 μm
Radius of the Fiber	a	190 μm
Activation Energy	Q	310 KJ/Mole
Norton Exponent of Creep	n, m	5
Matrix Young Modulus	E_m	70 GPa
Fiber Young Modulus	E_f	402 GPa
Matrix Poisson ratio	ν_m	0.34
Fiber Poisson ratio	ν_f	0.25
Matrix Thermal Expansion Coefficient	α_m	$10.6 * 10^{-6}$/C°
Fiber Thermal Expansion Coefficient	α_f	$8.9 * 10^{-6}$/C°
Initial Elastic Stress		
Yield of the Matrix	σ_0	1050 MPa
Yield Strength of the fiber	σ_f	380 MPa

In order to understand the effect of thermal history and heat treatment cycle, the concentric cylinder model introduced above was used to simulate the processes involved during the experiment. The data from Kim [18] for the matrix of the intermetallic phase was used to derive the parameters for the eq. 2. Although there may have been a difference in the grain size and composition, it provides a close approximation to the inelastic property of the matrix. The result of the simulation after the heating cycles and at room temperature is presented in Figs. 3–4.

A complete heat treatment cycle is presented in Fig. 3. Similar results were presented by Chandra *et al.* [8] who used a finite element method to model the residual stresses caused by different processing techniques in Ti-24Al-11Nb reinforced with SCS-6 fibers. Similar numerical techniques were developed by Jahaniani [19]. But none of these models can predict the effect of heat treatment duration. Simulation of the heat-treatment at different temperatures shows that the residual stress is highest for the shortest time of heat treatment (Fig. 4). A lower residual stress value was found for the case of 593 C° for 500 hours of heat treatment. These results show that, as the material cools down from its processing temperature, its von Mises stress initially varies elastically, and continues in the elastic-plastic regime until the material reaches room temperature. If the temperature is increased, the stress decreases linearly up to a certain "optimal" temperature, where the stress is very small, and then begins to increase linearly. After reaching the heat treatment temperature, the cooling path follows a path similar to the original cooling process, reaching an elastic-plastic region. This phenomenon is in

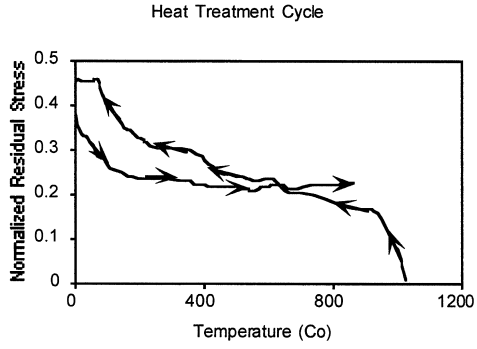

Figure 3. Model prediction for the residual stress evolution after heat treatment at different temperatures.

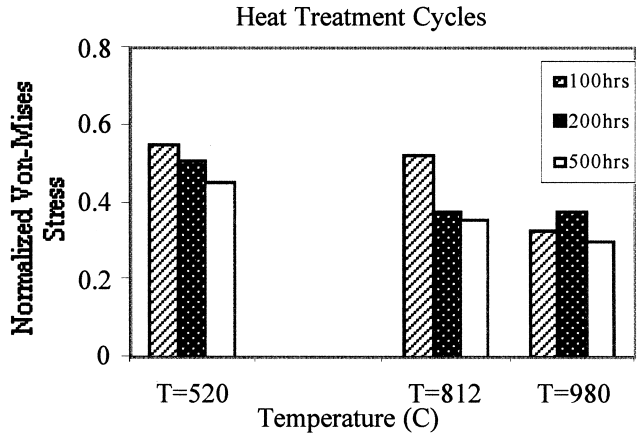

Figure 4. Model prediction for residual stress of Sapphikon fibers during the heat treatment cycle.

agreement with the results collected in this paper. An increase in the time of heat treatment has a similar effect on the material that an increase in temperature has: the stress increases or decreases depending on the temperature relative to the optimal temperature, with additional time. This could be explained based on the recovery of stresses accommodated by creep processes according to eq. 2. The von Mises stress for the composite reinforced alumina is initially large at 593°C, decreases and begins to increase at 815°C. The optimal temperature is between 593°C and 815°C. For the Sapphikon reinforced composite, the von Mises stress starts at a high level, decreases with an increase in temperature until it reaches the optimal temperature, which is between 815°C and 982°C, then the stress begins to increase. This behavior is also evident in the prediction of the model as shown in Fig. 4 specifically for 200 hours of heat-treatment. For low heat-treatment time of 100 this behavior is not evident. For much higher time of heat-treatment (500 hours) the residual stress keeps increasing as the temperature is increased.

Summary and Conclusion

X-ray technique was used together with micro-indentation to measure average residual stresses and local hardness of a titanium aluminide based composite (Ti-47Al-2Ta) material fabricated by HIP'ing. The measured values of residual stresses were compared to a composite cylinder model, which incorporates the effect of creep. In general, the residual stresses decrease as heat treatment temperature increases, reaching a minimum and then increases as the temperature and/or time of treatment increases. The fiber-matrix interface shows no degradation during the heat treating process. Such results compare well with the composite cylinder model once the effect of creep is incorporated. The analysis of the micro-hardness results indicate an increase in hardness in regions close to the fiber-matrix interface at 593°C but a softening occurs at 815°C. This behavior is reversed once the heat treatment temperature is increased to 982°C. At this temperature, the hardness at the interface increases for prolonged heat treatment duration. The model presented here shows that variation of thermal history will not fully eliminate the residual stresses, however, it is clear that recovery or high temperature creep contributes to the reduction of residual stresses.

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References

1. S. M. Arnold, V. K. Ayra, and M. E. Melis, NASA Technical Memorandum No. 103204 (1990).
2. S. M. Arnold and T. E. Wilt, NASA Technical Memorandum No. 105438 (1992).
3. P. Bartolotta, J. Barrett, T. Kelly, and R. Smashey, JOM. 49(5), 48–50 (1997).
4. J. Montgomery, M. G. Wells, B. Roopchand, and J. W. Ogilvy, JOM. 49(5), 45–47 (1997).
5. N. Hoo and P. S. Goodwin, JOM. 48(7), 40–41.
6. D. Newkirk and J. W. Dixon, in Proceedings of the 5th Symposium on High-Temperature Ordered Intermetallic Alloys, Boston, Nov. 30–Dec. 3, 1992, vol. 288, pp. 1063–1068, Materials Research Society, Pittsburgh (1993).
7. N. Hoo and P. S. Goodwin, JOM. 48(7), 40–41.
8. N. Chandra, C. R. Ananth, and H. Garmestani, J Compos. Technol. Res. 16, 37 (1994).
9. B. D. Cullity, Elements of X-Ray Diffraction, 2nd ed., pp. 447–478, Addison-Wesley, Boston (1978).
10. I. C. Noyan and J. B. Cohen, Residual Stress Measurement by Diffraction and Interpretation, p. 117–163, Springer-Verlag, New York (1987).
11. M. Ebrahimi, M.S. Thesis, The Florida State University (1993).
12. P. Rangaswamy and N. Jayaraman, J. Compos. Technol. Res. 17, 43 (1995).
13. P. Rangaswamy, W. C. Revelos, and N. Jayaraman, J. Compos. Technol. Res. 16, 47 (1995).
14. H. Dîlle and J. B. Cohen, Metall. Trans. A. 11A, 831 (1980).
15. M. McLean, Comps. Sci. Technol. 23, 37 (1985).
16. B. Fabeny and W. A. Curtin, Acta Mater. 44, 3439 (1996).
17. M. W. Hyer, Stress Analysis of Fiber-Reinforced Composite Materials, p. 110, McGraw-Hill, (1998).
18. Y. W. Kim, R. Wagner, and M. Yamaguchi, Gamma Titanium Aluminides, p. 967, The Minerals, Metals & Materials Society, (1995).
19. S. Jahaniani and M. Sabbaghian, ASME Journal of Pressure Vessel Technology 112, 85 (1990).